NONELASTIC STRAIN RECOVERY OF COPPER SINGLE CRYSTALS AFTER SMALL PRESTRAINS

In a recent paper[1] the details of measuring the total nonelastic strain recovery (γ_{pr}) during unloading of a zinc crystal (prestrained an amount γ_p) have been described. γ_{pr} was considered to be primarily related to the dislocation density of the crystal. That is,

$$\gamma_{pr} = \gamma_B + \rho bL \tag{1}$$

where γ_B is the strain recovery due to unbowing of dislocations, ρ the density of dislocations partaking in strain recovery, b their Burgers vector, and L the average distance they move during unloading. It was further assumed $d\gamma_B / d\gamma_p \simeq 0$ so that

$$\frac{d\gamma_{pr}}{d\gamma_p} \simeq b\,L\,\frac{d\rho}{d\gamma_p} \tag{2}$$

and

$$\frac{d\gamma_{pr}}{d\tau_p} \simeq b\,L\,\frac{d\rho}{d\tau_p} \tag{3}$$

where τ_p is the stress level reached in producing the strain γ_p. Relations (2) and (3) are valid if the specimen exhibits linear work-hardening and if L is much less sensitive to τ_p or γ_p than ρ . It is the purpose of this letter, to point out that such a description of γ_{pr} is in fair agreement with the stress and strain dependence of the dislocation densities of 99.999% pure copper crystals determined after small prestrains by Young[2] and Averbach and Rosenfield[3]. The authors prefer not to use the stress or strain versus dislocation density relations reported by Livingston[4] because his data cannot be extrapolated to the low stress and strain regions being considered in this communication.

Cylindrical single crystals 0.5 in. in dia. and approximately 7.5 in. in length were grown from 99.999% pure ASARCO copper. The crystals were grown in a split graphite crucible under a vacuum of 10^{-3} Torr using a modified Bridgman technique. Each crystal was cut in half, and slightly oversized cylindrical aluminum grips were fixed to both specimens with Shell Epon Adhesive VI. The crystals were chemically polished prior to testing. The orientations of some of the specimens tested, including a 0.01 atomic % aluminum alloy crystal (Cm-AL) are shown in Fig. 1. The method for determining γ_{pr} as a function of the maximum prior stress (τ_p) or plastic prestrain (γ_p) was almost identical to that described by Roberts and Brown[1]; the only difference was that tensile, not shear, tests were performed. The tensile capacitance gauge described by Hartman, et al.[5] was employed. Figs. 2 and 3 show γ_{pr} versus γ_p and γ_{pr} versus τ_p, respectively, for the crystals depicted in Fig. 1. Other pure copper crystals were tested and their results were similar to those reported here. All prestrains and measurements of γ_{pr} were carried out at room temperature (about 23°C). In Figs. 2 and 3 the vertical arrows represent the onset of stage II hardening. This was readily determined since the stress-plastic strain curve was constructed from the accumulated incremental prestrain tests at room temperatures. A very limited easy glide region was observed and this is attributed to the large size crystals tested[6] and their orientations. It should be noted that the crystal size employed was about the same as that used by Young[2], Averbach and Rosenfield[3], and Rosenfield and Averbach[7]. The large amount of scatter in the data is attributed to both the fact that the strain calibration is accurate between + 8% to -3% and that the determination of γ_{pr} involves drawing a tangent line to the unloading curve[1]. It should be pointed out that the absolute values of γ_{pr} noted in the current study are compatible with the very few results reported by Rosenfield and Averbach[7].

The interesting feature of Figs. 2 and 3 is that γ_{pr} increases about linearly with either τ_p or γ_p, within the experimental accuracy, except at very low stress and strains. From Fig. 2 $d\gamma_{pr}/d\gamma_p$ is approximately 3.5, 1.3 and 2.7 ($\times 10^{-3}$) for crystals CM-AL, CM-1T and CM-4T, respectively. Young[2] reports $d\rho/d\gamma_p \simeq 2.8 \times 10^{8}$ cm.$^{-2}$ for $8 \times 10^{-3} < \gamma_p < 6 \times 10^{-2}$, whereas Averbach and Cohen[3] suggest $d\rho/d\gamma_p \simeq 4 \times 10^{12}$ cm.$^{-2}$ for $0 < \gamma_p < 10^{-3}$. Using Young's[2] value for $d\rho/d\gamma_p$, b for Cu being 2.56×10^{-8} cm., and employing relation (2), one finds L to be 4.9, 1.8 and 3.8 microns for crystals CM-AL, CM-1T and CM-4T, respectively. These values are not unreasonable in the light of the fact that Young[8] has actually observed backward motion of dislocations from 1 to 6 microns in similar copper crystals after unloading from a stress of 10 g/mm^2. Young[2] also noted that

$$\tau_p \simeq 2 \times 10^{-4} \ \rho^{0.85} \ \text{gm/mm}^2 \ \text{for} \ \rho > 2 \times 10^{6} \ \text{cm.}^{-2}$$

from which $d\rho/d\tau_p$ is $\simeq 6 \times 10^{-4}$ gm^{-1} if one assumes $\rho \simeq 2 \times 10^{6}$ cm.$^{-1}$ and also neglects the weak stress dependence of $d\rho/d\tau_p$ in this region. Since $d\gamma_{pr}/d\tau_p$ is approximately 2.5, 1.5 and 0.32 ($\times 10^{-7}$) mm^2/gm for crystals CM-AL, CM-1T and CM-4T, respectively, (see Fig. 3) and using the previously discussed value of $d\rho/d\tau_p$ in relation (3), one finds L to be 1.6, 1.1 and 0.2 microns for crystals CM-AL, CM-1T and CM-4T, respectively. The agreement with the values of L determined from the experimental strain-dislocation density relationship is considered rather good in the light of the assumptions employed in using relations (2) and (3).

If the initial dislocation density of the crystal is small, namely $\rho < 2 \times 10^{6}$ cm.$^{-2}$, then Young's data[2] suggests $d\rho/d\tau_p \propto \tau^3$, which may account for the very rapid rise of γ_{pr} with γ_p for crystal CM-1T (Fig. 3) at stresses below 70 gm/mm^2.

The authors do not believe the current results assist in any way in differentiating between the various theories for stage II hardening in copper as briefly summarized recently by Wiedersich[9]. The authors simply wish to point out that the magnitude of the nonelastic strain recovery of copper after small prestrains appears to be primarily related to the dislocation density of the crystal at the time of stressing and is not appreciably sensitive to the distance over which they can move. The fact that each specimen exhibits a slightly different magnitude of γ_{pr} for constant τ_p or γ_p is not completely understood. From the present discussion, γ_{pr} should be sensitive to both the initial dislocation density of the crystal as well as its ability to accumulate new dislocations per unit strain. Structural variations from one crystal to another could easily account for such differences. For example, it is interesting to note that the present results are in agreement with the conclusions drawn by Young and Savage(10) concerning the perfection of copper crystals prepared by the Bridgman technique. They found that for more impure crystals, other preparation factors being equal, the crystals were less perfect and contained a large dislocation density. This may explain why the curve for crystal CM-AL (Fig. 2) lies above those for the purer crystals. Young and Savage(10) also noted that 99.999% pure copper crystals were more perfect, the further removed the specimen axis was from a [111] zone. Since crystal CM-1T is further removed from the [111] zone than CM-4T, (Fig. 1), this orientation effect may be related to the fact that γ_{pr} at constant τ_p or γ_p for crystal CM-1T tends to fall below the data points for crystal CM-4T (Figs. 2 and 3).

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FIGURE CAPTIONS

Figure 1. Portion of a standard [001] stereographic net showing the pole of the specimen axis for each of the crystals studied and reported in detail here.

Figure 2. γ_{pr} versus γ_p for various specimens.

Figure 3. γ_{pr} versus τ_p for the same specimens described in Figs. 1 and 2.